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Physically based material model for evolution of stress–strain behavior of heat treatable aluminum alloys during solution heat treatment

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ABSTRACT

A mathematical model based on Kocks–Mecking–Bergstrom model, has been proposed to predict the flow behavior of age hardenable aluminum alloys, under different conditions of solution heat treatment and hot deformation. Considering the published literature, most researchers have taken into account the precipitation and solution strengthening contribution to the flow stress by a constant and some others have ignored these effects. So these available descriptions cannot be applicable directly to different conditions of solution heat treatment. In order to enable these constitutive descriptions to take into account the effects of soaking time and temperature, we introduce in this research a relative volume fraction of precipitation into the flow stress by using the appropriate relationships. The GA-based optimization technique is used to evaluate the material constants within the equations from the uniaxial tensile test data of AA6082 reported by previous researchers.

1. Introduction

There is a number of approaches to model work hardening behavior by applying a physically based material model. Kocks and Mecking [3], Bergstrom [1] and Bergstrom and Hallen [2] have proposed a model based on dislocation evolution (K–M–B model), to describe the plastic flow [1–3]. In this model, they have considered the hardening process as a competition between dislocation accumulation and the loss of dislocation line length (i.e., dynamic recovery). Although the ability of this kind of constitutive relationship, based on micro-structural evolution during forming, has been proved in many experiments [1–9], but the main difficulty in using these relationships is how to accurately determine their constants. Moreover, when a complex problem such as the optimization of the heat treatment for precipitation hardening or solution heat treatment of aluminum alloys is considered, the available theoretical descriptions are less secure, because in those descriptions the influence of these parameters (precipitation and solution hardening) are usually considered as a constant or sometimes those are ignored. Obviously, these effects are not constant and depend on thermomechanical history of the alloys and how the chemical elements of the alloy are distributed in the matrix.

Heat treatable aluminum alloys (e.g. 6xxx) in the aged condition have relatively a high strength and low ductility that refers to the resistance of precipitates to dislocation motion [10,11]. So, at this condition the workability of alloys is relatively low. The usual solution to overcome this problem is solution heat treatment which refers to the treatment where the second phase particles become unstable and dissolve into the matrix phase. Under this condition the resistance to deformation is reduced and ductility is raised. Although several works concerning the hot flow behavior of aluminum has been carried out so far [4,7,8,11], but to the knowledge of the authors there is not a mechanism based model to predict the flow behavior of heat treatable aluminum alloys under hot deformation conditions by consideration of soaking time and temperature, work hardening and work softening at different conditions of deformation.

In order to develop such a model, in this research, the K–M–B model, is extended to handle the variation of dislocation density during hot working, and precipitation and solid solution hardening contribution to the flow stress by considering the kinetics of precipitation dissolution during soaking and total concentration of alloying elements. The constants in the developed constitutive equation are determined using the genetic algorithm (GA) based optimization technique and the experimental data of AA6082 reported by Garrett et al. [11]. The difficulty of choosing suitable starting values for the constants in the traditional optimization techniques are completely overcome since the GA technique provides a better chance to converge to the global minimum [12].

2. Mathematical modeling

Determination of flow stress during deformation is a key requirement of the modeling of metal forming processes [8]. This may be carried out by considering the barriers to the motion of

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dislocations. The common barriers are the forest dislocation and friction stress of the metal [2,9]. The contributions of these parameters to flow stress can be evaluated as below.

2.1. Contribution of forest dislocation

A generally accepted assumption for the influence of dislocation density on flow stress is [2,4,9]

$$\sigma_w = \alpha G b \sqrt{\rho} \tag{1}$$

where ρ is the forest dislocation density, G the shear modulus, b the magnitude of the Burger's vector, and α a constant of order of unity depending in part on the strength of the dislocation/dislocation interaction.

In Kocks–Mecking–Bergstrom (K–M–B) model, the dislocation density, ρ, is a result of the balance between the dislocations stored during work hardening (hardening term) and the dislocations annihilation during DRV (recovery term). The effect of these competing phenomena has been described by [1–3,5,6]

$$\frac{d\rho}{d\varepsilon^p} = k_1\sqrt{\rho} - k_2\rho \tag{2}$$

where ε^p is the plastic component of strain. The first term, $k_1\sqrt{\rho}$, describes the storage rate, and the second term, $k_2\rho$, is associated with dynamic recovery. In the Bergstrom model, k_1 is the rate of immobilization or annihilation of mobile dislocations and k_2 the probability of remobilization or annihilation of immobile dislocations [1]. The remobilization is a thermally activated process, and at high temperatures it is based on vacancy climb [2,6,7]

$$k_2 = k_0 + k(\dot{\varepsilon}, T) \tag{3}$$

where

$$k = A(\dot{\varepsilon})^{\frac{-1}{3}} \exp\left(\frac{-Q_m}{3RT}\right) \tag{4}$$

k_0 and A are constants, Q_m is the vacancy migration energy, and R and T have their usual meanings.

Eq. (2) can be analytically integrated using the boundary condition $\varepsilon^p = 0$ at $\rho = \rho_0$, and then it is combined with Eq. (1) to present the flow behavior of the alloy as

$$\sigma_w = \alpha G b\left[\sqrt{\rho_0} + \left(\frac{k_1}{k_2} - \sqrt{\rho_0}\right)\left(1 - \exp\left(\frac{-1}{2}k_2\varepsilon^p\right)\right)\right] \tag{5}$$

where ρ_0 is the initial dislocation density. At a constant temperature and strain rate this equation is the same as the Voce hardening function [7,8].

2.2. Contribution of friction stress

The flow stress given by Eqs. (1) and (5) relates only to the impediment to dislocation motion provided by other dislocations. In most materials, there are other contributions, like lattice resistance, precipitation hardening, solution hardening, and some grain size effect to plastic resistance. In some cases these are added to the contributions discussed above [2–4,9]. So, the flow stress, σ, will be determined as

$$\sigma = \sigma_f + \sigma_w \tag{6}$$

where σ_f is the extrapolated stress to zero dislocation density (strain-independent stress "friction stress"), corresponding to short-rang interactions. Thermal vibrations can assist the friction stress to overcome these interactions. For the case of pure f.c.c. metals, the friction stress can be neglected, as shown by experiments [3], but, as exhibited in this work, in the case of alloys the magnitude of friction stress is noticeable and cannot be ignored. The second term, σ_w, is due to long-rang interactions with the dislocation substructure. For the case of strain-independent friction stress one can assume [7]

$$\sigma_f = \sigma^*(\dot{\varepsilon}, T) + \sigma_0(T) \tag{7}$$

where $\sigma_0(T)$ is the strain and strain rate independent component of flow stress and $\sigma^*(\dot{\varepsilon}, T)$ is the dynamic stress depending on the stain rate and temperature.

Dynamic stress is usually very small for f.c.c. alloys and as temperature rises, this part of the friction stress quickly drops to zero [9,7]. Therefore, the dynamic stress is neglected in the current study. The expression for the strain and strain rate independent component of friction stress is assumed to be [9,13]

$$\sigma_0 = \sigma_{ss} + \sigma_p + \sigma_i \tag{8}$$

where σ_{ss} is the solution hardening component, σ_p the precipitation hardening component, and σ_i the intrinsic strength of the pure aluminum.

At room temperature, the effect of σ_{ss} and σ_p on flow stress can be determined as below.

2.2.1. Solid solution hardening

The effect of solid solution hardening at room temperature, σ_{ss}, on friction stress is related to relative volume fraction of precipitates, f_r, in matrix as [14,15]

$$\sigma_{ss} = \sigma_{0ss}(1 - \alpha_2 f_r)^{\frac{2}{3}} \tag{9}$$

where σ_{0ss} is the solid solution contribution in the as-quenched condition at room temperature and α_2 is the fraction of the as-quenched solute concentrations (assuming a quasi-binary system) depleted from the matrix when f_r approach unity at the peak-age condition.

In age hardenable aluminum alloys, the elements such as Mg, Si and Cu give rise to considerable solid solution strengthening. Provided that the contribution from each element is additive, the solid solution potential of the alloy, in the as – quenched condition, σ_{0ss}, can be expressed as [13]

$$\sigma_{0ss} = \Sigma a_j C_j^{\frac{2}{3}} \tag{10}$$

where C_j is the total concentration of a specific alloying element in the alloy and a_j is the corresponding scaling factor. Thus, for a given alloy we can assume σ_{0ss} as a constant.

2.2.2. Precipitation hardening

For simplicity, particle shearing can be assumed as the dominating strengthening mechanism. Under such condition, the net precipitation-strength enhancement after an arbitrary reheating cycle at room temperature, is given as [16];

$$\sigma_p = H\left[(f)^{\frac{1}{2}}(r)^{\frac{1}{2}}\right] \tag{11}$$

where H is a constant, r and f are precipitation radius and volume fraction, respectively. Similarly, for an initial condition such as the peak-age condition where $f = f_0$ and $r = r_0$ one may write by analogy

$$\sigma_{p,m} = H\left[(f_0)^{\frac{1}{2}}(r_0)^{\frac{1}{2}}\right] \tag{12}$$

$$\alpha_3 = \frac{\sigma_p}{\sigma_{p,m}} = \left(\frac{f}{f_0}\right)^{\frac{1}{2}}\left(\frac{r}{r_0}\right)^{\frac{1}{2}} = \left(\frac{f}{f_0}\right)^{\frac{2}{3}} = \left(\frac{r}{r_0}\right)^2 \tag{13}$$

where α_3 is referred to as the dimensionless strength parameter.

Therefore

$$\sigma_p = \sigma_{p,m}(f_r)^{\frac{2}{3}} \tag{14}$$